Extended Abstracts of the 1992 International Conference on Solid State Devices and Materials, Tsukuba, 1992, pp. 446-448

Sensitivities of Laser/Microwave- and Conventional-DLTS for Defects in CZ Silicon

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The sensitivity of a noncontact and nondestructive laser/microwave (LM) DLTS for defect characterization was confirmed with heat-treated Czochralski (CZ) silicon crystals. The carrier trap levels (E_T) in samples subjected to a three-step intrinsic gettering (IG) thermal process were clearly obtained with the LM-DLTS even though these samples showed no significant oxygen precipitation. The sensitivities of LM- and conventional (C)-DLTS for defect diagnosis were compared each other, and the LM-DLTS is concluded to be much more sensitive to bulk microdefects in silicon crystals. The detection limits of LM- and C-DLTS in defect density are lower than $1x10^7cm^{-3}$ and around $1x10^{10}cm^{-3}$, respectively.

Introduction

In the field of ULSI, Czochralski (CZ) silicon wafers are the main substrates for device production. Various defects and impurities can be introduced into silicon substrates during device fabrication processes, and these defects greatly affect the device performance. Since ULSI devices are greatly influenced by the quality of substrate, high sensitive diagnostic techniques are required to characterize these defects. Moreover, from the material characterization point of view, especially from the in-line monitoring of view, noncontact and non-destructive methods which require no specific sample preparation are desired since those are free from introducing additional defects and impurities.

Recently, a noncontact and nondestructive laser/microwave (LM)-DLTS technique [1,2] has been applied to characterize crystal defects near the wafer surface and/or in the substrate bulk [3,4]. However, quantitative discussions for the LM-DLTS in the sensitivity has not been performed yet. In this study, carrier trap levels (E_T) due to bulk crystallographic defects in CZ silicon wafers, which were subjected to a three-step intrinsic gettering (IG) thermal process, are studied by means of two different DLTS techniques, *i.e.*, LM- and conventional (C)-DLTS.

Experimental

CZ silicon wafers used in this study are *n*(100) 150-mm-diam. (20~30ohm-cm) with different initial oxygen ($[Oi]_0$: 26~33ppma) and carbon ($[Cs]_0$: <0.02 or 0.2ppma) concentrations [5,6] as shown in Table I. These crystals were grown under the identical growth conditions except for $[Oi]_0$ and $[Cs]_0$ control. Since defect generation during subsequent thermal cycles is greatly affected by the thermal environment in crystal growth as well, the body length of all the ingots was controlled to be 600mm to minimize the variation in thermal history of crystals. Wafers were prepared from a middle portion; *i.e.*, 200~300mm from the seed end of the ingots. These samples were subjected to a three-step IG heat treatment (1st: 1100°C/4h, 2nd: 750°C/64h, and 3rd: 1000°C/16h) in a mixed gas (95% N_2 + 5% O_2), but were subjected to neither extrinsic gettering treatment nor pre-annealing such as for donor annihilation to avoid any other effects on defect generation.

After each step, all the samples were measured with LM-DLTS (50~200°C) and C-DLTS (-180~100°C) methods. The E_T was measured with LM-DLTS technique using LIFETECH-88® (SEMITEX Co., Ltd.) after each step. The measurement is based on analyzing the decay of minority-carriers generated by irradiation with a pulsed laser beam (λ=910nm). In general, the effective recombination lifetime, τ_{eff}, consists of the surface (τ_S) and bulk (τ_b) components; that is,

$$1 / \tau_{eff} = 1 / \tau_S + 1 / \tau_b \qquad [1],$$

Therefore, τ_{eff} was measured for wafers with thermal oxide grown during each heat treatment to minimize the τ_S. In the case of *n*-type silicon, the relation between the rate window ($W = 1 / \tau(T_m)$) and an E_T can be expressed as

$$\ln(WT) = (E_T - E_C) / kT_m \qquad [2],$$

where E_C is the conduction band energy in silicon. The E_T was obtained from a slop of the Arrhenius plot, $\ln(WT)$ vs $1/T_m$. No specific sample preparation was performed for LM-DLTS measurements.

For C-DLTS measurements, however, specimens were prepared in order to investigate the bulk region as follows; (i) thermally-grown oxide was removed with buffered HF solution, (ii) about 50μm surface layer was etched-off with HF and HNO_3 mixed solution to remove the denuded zone (DZ) layer, and then (iii) evaporated Au for Schottky contact (20nm) and Al for ohmic contact (80nm).

For quantitative discussion, the bulk defect densities (D_d) after the third step were measured with light scattering tomography (LST) [7,8].

Results and Discussion

During the thermal process, oxygen precipitation occurs resulting in [Oi] reduction due to the formation of oxygen-related defects. The oxygen reduction (Δ[Oi]) in the samples after each step are listed in Table I.

Table I $[Oi]_0$ and Δ[Oi] (ppma) after each step for samples used in this study.

Sample	as-grown	1st	2nd	3rd
A*	27.4	0.3	0.3	1.5
B**	26.4	0.0	0.0	9.2
C*	29.7	0.1	0.3	1.3
D**	29.8	0.5	0.5	22.9
E*	32.3	0.0	0.0	0.4
F**	33.0	0.1	0.5	29.8

* [Cs]<0.02ppma, **[Cs]=0.2ppma

After the first and second steps, no significant Δ[Oi] is observed in the samples. After the third step, however, Cs-rich samples (B, D, and F) show much greater Δ[Oi] in contrast with Cs-lean samples (A, C, and E). It has been well known that carbon impurity enhances oxygen precipitation in silicon[9,10]. The greater Δ[Oi] after the third step can be attributed to greater $[Cs]_0$, and this result suggests that oxygen-related microdefects were generated in Cs-rich samples during the second step even though these samples showed no significant oxygen reduction. In fact, these Cs-rich samples showed lower τ_b after the second step (20~700μs) comparing with those in samples after the first step (2000~3000μs). The τ_b redu-ction indicates the generation of carrier trap centers in these samples originated from bulk microdefects.

Table II summarizes the E_T obtained with both LM- and C-DLTS, and bulk defect density measured with LST after the third step. Many kind of crystal defects were exist in these oxygen-precipitated samples; *i.e.*, oxygen precipitates, dislocation loops, stacking faults, and point defects associated with silicon self-interstitials generated during the oxygen precipitation process. Hwang and Schroder have reported [11] that oxygen precipitates are mainly responsible for the degradation of recombination lifetime, and the carrier recombination at oxygen precipitates takes place through the interface between the precipitates and matrix silicon. Since the LM-DLTS is analyzing the temperature dependence of recombination lifetime, the obtained trap levels in the present experiment might be attributed to oxygen precipitates and/or oxygen-related microdefects.

Table II Activation energies (E_a) obtained with C- and LM-DLTS after each step, and bulk defect density (D_d) after the third step.

X'tal	1st step (eV)		2nd step (eV)		3rd step (eV)		D_d
	C	LM	C	LM	C	LM	(cm^{-3})
A	ND	ND	ND	ND	ND	0.1	<1E07
B	ND	ND	ND	ND	ND	0.2	4.3E09
C	ND	ND	ND	ND	ND	0.1	9.0E07
D	ND	ND	ND	0.2	0.3	0.2	1.7E10
E	ND	ND	ND	ND	ND	0.1	2.8E08
F	ND	ND	ND	0.2	0.3	0.2	2.1E10

The LM-DLTS signals were recognized for all the samples after the third step; however, C-DLTS gave rise to E_T only for samples D and F which were higher than $10^{10}cm^{-3}$ after the third step in D_d. The high sensitivity of LM-DLTS is remarkably reco-gnized for samples D and F after the second step. For example, the trap level in sample F is clearly obtained with this technique even though this sample showed no significant oxygen preci-pitation as shown in Fig.1.

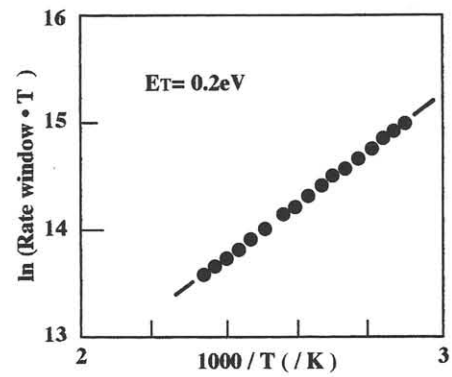

Fig.1 A representative Arrhenius plot obtained with LM-DLTS for wafer F after the second step.

The detection limits of LM- and C-DLTS in defect density are lower than $1x10^7cm^{-3}$ and around $1x10^{10}cm^{-3}$, as shown in Fig.2, respectively. In the LM-DLTS technique, measured volume (V_L), which gives DLTS signal, is limited by the carrier diffusion length (L); that is, $L=(D\tau)^{1/2}$, where D is diffusion coeffi-cient and τ is carrier lifetime. Therefore, the V_L is calculated as $V_L = 4\pi L^3 / 3$. On the other hand, the C-DLTS technique collects signals from a volume (V_C), which is limited both by an area of Schottky contact (S) and depth of the depletion layer (d), therefore the V_C is given by $V_C = Sd$. In the present study, the S was $3.14x10^{-2}cm^{-3}$ and d was ~10μm. Taking into account this difference in the measured volume, minimum number of crystal defects which give LM- and C-DLTS signals were estimated to be ~$1x10^3$ and ~$3x10^5$, respectively.

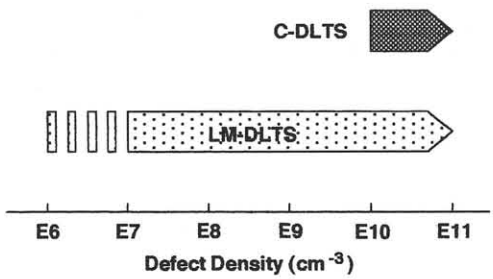

Fig.2 Sensitivity of C- and LM-DLTS for defect characterization.

The E_T for samples D and F after the third step, were obtained about 0.2eV with LM-DLTS, and about 0.3eV with C-DLTS. Both in LM- and C-DLTS, the time constant of transient signals is proportional to $\exp[(E_C-E_T)/kT]$, and usually the E_C is treated to be constant value. However, the temperature dependence of energy band gap (E_g) should be considered to compare the obtained E_T values with LM- and C-DLTS, because of their wide difference in the measuring temperature range. The E_g at temperature T is given by [12]

$$E_g(T) = 1.17 - (4.73x10^{-4})T^2 / (T+636)\ eV \quad [3],$$

therefore, the energy difference in the E_g is calculated to be ~0.1eV at 200 and 500K. This fact indicates that the E_T obtained with LM-DLTS is shallower than that of obtained with C-DLTS, even though the E_T is originated from the same source.

Conclusion

The sensitivity for the defect characterization of a noncontact and nondestructive LM-DLTS technique was confirmed in this work. This technique is much sensitive than C-DLTS to crystal defects in silicon, and the E_T can be obtained with LM-DLTS even though the samples show no significant oxygen precipitation. Considering advanced ULSI device fabrication processes which require extremely clean environment and high quality silicon substrates, one may expect that the LM-DLTS can be a powerful in-line diagnostic tool because of its contactlessness, nondestructiveness, and high sensitivity.

Acknowledgements

The authors would like to acknowledge Dr. A. Buczkowski (NCSU) for his helpful discussion, and Mr. K. Iba (SEMITEX Co., LTD.) for his help in LM-DLTS measurements.

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